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COMMENTS ON THE MEASUREMENT OF DISLOCATION MOBILITY AND THE DRAG DUE TO PHONONS AND ELECTRONS

by

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ABSTRACT

Experimental methods for the measurement of intrinsic interactions between moving dislocations and the crystal lattice are considered. It is emphasized that the stress pulse method is applicable at stress levels greater than about twice the static flow stress, while internal friction experiments may be used to explore the interaction at very low stress levels and small dislocation velocities. Recent results of low temperature stress pulse measurements in Cu are presented. The interactions deduced from measurements between 4.2°K and 400°K in some FCC metals are compared to theoretical predictions. Suggestions are made for future theoretical and experimental work on unresolved aspects of the intrinsic interactions.

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I. INTRODUCTION

The now classic work of Johnston and Gilman[1] in which dislocation velocities were measured as a function of stress in LiF lead the way to numerous measurements of the dynamics of dislocations in crystals. These measurements give information which is useful for predicting the deformation response of crystals, and for studying the origin and nature of the forces which impede dislocation motion. This paper examines the techniques used in experimental methods for the measurement of dynamic dislocation behavior. These methods fall into two categories: (i) indirect methods in which macroscopic parameters such as strain rate and stress or the energy loss in an ultrasonic stress wave are determined and (ii) direct methods in which dislocation velocities are deduced from dislocation displacement measurements in stress pulse tests or from observations of the time dependence of dislocation motion.

Experimental conditions for the determination of the strength of the interaction between a moving dislocation and an otherwise perfect lattice using the direct method are set forth. Theories of the temperature dependence of the interaction are discussed and compared to experimental measurements from $4.2^{o}K$ to $400^{o}K$. Unresolved aspects of the interaction between a moving dislocation and lattice phonons and conduction electrons are discussed.

II. INDIRECT AND DIRECT METHODS

Two indirect methods have been used for the measurement of the dynamic properties of dislocations: (i) measurements of internal friction and modulus decrement, and (ii) measurement of the strain rate vs. stress behavior at strain rates above about $10^2 sec^{-1}$.

Dislocation displacements are of the order of atomic dimensions in the internal friction measurements, and dislocation velocities vary up to about $\pm$ 1cm/sec about a mean value of zero. The Granato-Lücke[2] string model is usually used in the interpretation of the data.

Considerably larger dislocation displacements and velocities are involved in the strain rate vs. stress experiments. In this type of test only the product of mobile dislocation density and average dislocation velocity may be determined. Hence a knowledge of the mobile dislocation density is needed to determine an average dislocation velocity uniquely. A reliable measurement of the mobile dislocation density cannot be made, so that reliable estimates of dislocation velocities are difficult to obtain.

Since dislocation displacements are observed in the direct method, a theoretical model connecting dislocation dynamics with the measured parameters is not required. It must be recognized, however, that only an average velocity can be determined with this method. When the stress pulse technique is used and dislocation positions are determined before and after a rectangular stress pulse is applied, an average velocity given by the displacement divided by the pulse duration is found. When the dislocations are continuously observed during their motion, as in a transmission electron microscope, velocities are determined which are averages over the time period set by a motion picture framing rate or are averages over the time period set by a vidicon or motion picture framing rate or are used to observe the dislocation. Deviations from the average velocity which might take place as the displacement varies over atomic dimensions therefore cannot be detected. It is generally assumed that a terminal dislocation velocity is determined in the direct method and that the driving force supplied by the applied stress is equal to the retarding or drag force at that velocity.

We may divide the forces which impede dislocation motion into two categories, intrinsic and extrinsic. Intrinsic forces act on a dislocation in an otherwise perfect lattice while extrinsic forces are due to dislocation interactions with other lattice defects. Both intrinsic and extrinsic forces will act on dislocations under actual experimental conditions, since all crystals contain defects such as surfaces, vacancies, and impurities. Inertial forces are unimportant in most of the direct experiments.[3]

III. MEASUREMENTS OF INTRINSIC INTERACTIONS

We consider here the problem of the measurement of intrinsic interactions in FCC crystals and on close-packed planes of HCP crystals. The motion of dislocations through an array of discrete obstacles has been treated theoretically by Kocks,[4] Frost and Ashby[5] and others. It has been predicted[5] that local interactions with obstacles which obstruct dislocation motion become unimportant compared to velocity dependent intrinsic interactions at stress levels above about two times the critical stress required to drive the dislocation through the obstacles. At intermediate stress levels both intrinsic and extrinsic obstacles are effective; the stress-velocity relationship is non-linear and the total drag is greater than that due to intrinsic forces alone. These calculations were verified experimentally in zinc where the obstacle to basal dislocation motion was a forest of second-order pyramidal dislocations.[6] We conclude from this that reliable direct measurements of the intrinsic interactions must be made at stress levels greater than two times the stress which causes dislocations to move macroscopic distances. In the internal friction experiments, the dislocations vibrate about their equilibrium position and are subjected to intrinsic forces. The stress levels are very low compared to the stresses required to overcome all of the extrinsic obstacles (the stronger extrinsic interactions fix the length of the vibrating segments and this effect is included in the theory). The intrinsic interactions may then be studied at either very low stress levels (internal friction measurements) or high stress levels (strain rate vs. stress and the direct method.)

Extrinsic interactions are reduced by (i) minimization of impurities and the total dislocation density of the test crystals, (ii) using "fresh" dislocations which are free of segregated impurities and contain a minimum number of jogs, (iii) minimizing or avoiding interactions of dislocations on the active slip system or attraction of dislocations to free surfaces (these interactions may increase the driving force on the leading dislocations and cause the intrinsic drag forces to be underestimated). One method is described below which has been found effective in reducing extrinsic interactions so that the strength of intrinsic interactions could be determined. This method utilizes torsional stress waves to produce single, short duration stress pulses. Dislocation displacements are observed either on a cross section normal to the cylindrical specimen axis (the cross section is subjected to a stress distribution corresponding to that of static torsion) or on the cylindrical surfaces of the specimen (where the maximum torsional stress acts).

IV. TORSION TESTING

The method used to generate torsion waves with a 2μsec rise time has been described elsewhere.[7] A torsional stress state is advantageous in that microsecond duration elastic stress pulses are non-dispersive in isotropic rods, in ⟨100⟩ axis rods of cubic crystals and in [0001] axis rods of hexagonal crystals. A small amount of dispersion occurs in anisotropic crystal rods with a ⟨111⟩ or ⟨110⟩ axis but this dispersion is not a problem in the experiments.[8] Isolated dislocations are introduced for study by scratching[9] or by the use of line focused laser pulses.[10] The torsion waves are coupled to the test crystal by a gluing agent which has a liquid to glass transition temperature very near the test temperature. Suitable agents which do not cause dislocation displacements in the cooling and heating cycles have been found for test temperatures between 44°K and 400°K. The bonding agent used at 44°K is sufficiently strong at 4.2°K to pass stress waves whose amplitude exceeds 10^7 dyn/cm^2. Below 44°K, differential thermal expansions cause sufficient stress to move dislocations near the bonded surface during the heating and cooling cycles. Therefore only the mobility of dislocations intersecting the cylindrical surfaces of the crystal is studied below 44°K.

The basal dislocations in zinc revealed in the Berg-Barrett topographs of Figures 1 and 2 are about 10μ below the surface. They were produced by scratching the (0001) surface with an alumina whisker

(75mg load). The crystals were annealed prior to scratching so that all of the grown-in basal dislocations within about 20μ of the surface were removed by climb to the surface. After scratching a set of dislocations is observed parallel to the scratch and extending about 70μ on each side of it. The dislocation spacing is about 10μ and their interaction at this spacing is weak due to the presence of the nearby free surface. Figure 1 shows the dislocation configuration after application of a torsional stress pulse (the pulse was applied within 1 hour of scratching and its duration was about 20μsec). The continuous scratch to the right of the figure shows dislocation displacements which are directly proportional to the radial distance from the center of the cylindrical crystal. The torsional stresses vary linearly with radius, and the entire surface experienced the same duration of loading so the dislocation displacement vs. radius curve indicates directly the shape of the dislocation velocity vs. stress curve. Dislocation displacements from the discontinuous scratch to the left of Figure 1 show the effect of pinning at the ends of each scratch segment. The effect of end pinning is strong in the low stress region near the crystal center and small in the high stress region where the displacements are essentially the same as those of the unpinned dislocations on the continuous scratch. Figure 2 shows that the pinning is strong enough to permit the dislocation to act as a Frank-Read source. Cross slip of some screw oriented segments is indicated.

The theories discussed below indicate that the dominant intrinsic interactions take place very near to the core region of the dislocation. We conclude from this that a free surface parallel to the slip plane and about 10μ away from the dislocation core will have a small effect on the intrinsic drag since its effect on the dislocation strains and displacements near the core is small. Dislocation displacements near an end surface of a cylindrical crystal in a torsion pulse test should then be representative of the displacement of isolated dislocations in the bulk crystal.

Recently we have measured the mobility of 30^o mixed dislocations intersecting the lateral surfaces of cylindrical copper crystals at 4.2^oK by means of the torsion technique.[+] Dislocation displacement as a function of torsional impulse for these tests is plotted in fig. 3. The displacement data indicates that extrinsic interactions were important at the lower levels of torsional impulse. We believe that interactions between scratch produced dislocations on different slip systems are the major extrinsic interactions in these tests.

V. THE MAGNITUDE OF DISLOCATION -PHONON AND DISLOCATION-ELECTRON INTERACTIONS IN METALS

In solid materials with a small Peierls' barrier in the dislocation slip planes, such as in the primary $\langle 111 \rangle$ planes of FCC metals and the basal plane of HCP metals, lattice vibrations (termed thermal phonons) and conduction electrons are postulated to be responsible for the primary sources of damping of the motion of moving dislocations in an otherwise perfect lattice. Even in metals with large Peierls' barriers, emission and absorption of phonons is thought to dominate the energy-absorption process from moving dislocations

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[+] To be reported elsewhere.

in the moderate to high stress region.

These mechanisms have been described extensively in the literature, at least in the case of elastic isotropy. External to the dislocation core, three fundamental processes have been postulated to determine the influence of phonons on the mobility of dislocations. They include (i) the interaction of the elastic strain field of a dislocation with thermal phonons, (ii) radiation of thermal energy from stress-induced oscillations of a dislocation, and (iii) influence of a discrete lattice in which the group velocity of thermal phonons varies as a decreasing function of their energy.

The first of the above three mechanisms can be separated further into two components, strain field scattering[12] and phonon viscosity.[13,14] (There is the additional small contribution from macroscopic thermoelastic damping.[15]) Each source of dislocation damping has a characteristic velocity and temperature dependence and is expected to dominate in a particular temperature, dislocation velocity region.[16] Both effects are insignificant at low temperatures. In the region of the Debye temperature, energy dissipation by phonon scattering exhibits a linear increase with temperature, while the phonon viscosity effect becomes temperature-independent.

The second of the above mechanisms is operative whenever a dislocation is accelerating. Three conditions under which this may occur are described below. The acceleration of a dislocation from some uniform velocity (including zero velocity) under the influence of an external stress field, either in the case of free movement through the lattice or of its motion between obstacles in the crystal, is accompanied by the radiation of elastic energy.[17] A dislocation moving through a crystal lattice is constantly accelerated and decelerated by the oscillating stress field of the thermal phonons. This interaction, or 'flutter' mechanism[18-20] as it is commonly known, gives rise to the radiation of thermal energy. In addition, for a moving dislocation, the Peierls' barrier presents an oscillating stress field, with attendant thermal energy radiation.[21] This barrier is especially important in BCC metals. As the Debye temperature is exceeded, the thermal phonon mean free path decreases to the order of atomic dimensions. In this limit a significant volume of the dislocation strain field cannot be coherently excited by a thermal phonon, and the flutter mechanism becomes unimportant.

The third mechanism mentioned above has been more recently proposed[22,23] where it has been shown that in a discrete lattice at $0^{o}K$ a moving dislocation radiates energy when its velocity exceeds the group velocity of thermal phonons moving in its direction of motion. The analysis has been extended to finite temperatures to include the effect of strain field scattering.[24]

Calculation of the magnitude of the interaction between conduction electrons and a moving dislocation in metals at low temperatures shows that a temperature independent dissipative force is exerted on the dislocation.[25] It is thought that the dissipative force has less relative significance at more elevated temperatures, where the phonon dissipative mechanisms increase in magnitude.

Considerable experimental evidence has been accumulated in measurement of the magnitude of dislocation-phonon and dislocation-electron interactions. In the temperature range 4.2°K to room temperature, dissipative forces, measured in direct mobility experiments where extrinsic effects were small, have been shown to increase linearly with dislocation velocity up to about ten percent of the shear wave velocity.[8, 10, 26-29] This is in accord with the linear velocity-stress relation predicted by the conduction electron and each thermal phonon damping mechanism. (The dissipative force is expressed as B(T)v, where B(T) is the damping coefficient at temperature T and v is the dislocation velocity.)

In addition, results of both the direct experiments referenced above and indirect measurements (e.g. References 30-32) from about 30°K to room temperature, have exhibited an increasing dissipation with temperature, in qualitative agreement with the increasing influence of thermal phonons with higher temperature. (There appears to be reasonable correlation of results from direct measurements with those indirect measurements in which a somewhat narrow peak in the attenuation versus frequency curve was recorded. It has been pointed out[33] that a wide peak may preclude treatment of the attenuation vs. frequency data on basis of the most simple Granato-Lücke model.) Much experimental evidence also indicates that as the Debye temperature (of the order of room temperature for many metals with close-packed structures) is approached, the rate of increase of the damping coeffient is lessened. In the only direct measurement made above room temperature,[8] the edge dislocation damping coefficient in copper was found to be independent of temperature between 296°K and 373°K. While no single mechanism has as yet been shown to be dominant over the extended range of temperature in which mobility measurements have been made, the trend in B at higher temperatures ($dB/dT \rightarrow 0$) is qualitatively supportive of phonon viscosity in this region. (It is interesting to note that four independent estimates of B in copper from internal friction data at room temperature[30, 32-34] all fall within $1\text{-}2 \times 10^{-4}$ cgs. In three of the four cases, additional data extended the possible limits of B to a somewhat wider range of values. This compared with $B \simeq 2 \times 10^{-4}$ cgs from direct measurements.)

Brailsford[16] has argued that the dominant Fourier components of a moving dislocations' displacement field dissipate energy through the process of strain field scattering, while the viscosity effect cannot be important. According to his calculation for strain field scattering the dislocation damping coefficient in copper at room temperature is given by $B(Cu) \simeq 3.9 \times 10^{-3}$ cgs, or more than one order of magnitude larger than that indicated by the accumulated data. Mason, who originally postulated that a moving dislocation dissipates energy through phonon viscosity, estimated the damping coefficient in copper to be $B(Cu) \simeq 2.5 \times 10^{-4}$ cgs at room temperature.[13] Mason's formula for B contained an inverse dependence on the square of an inner "cut-off" radius about the dislocation center within which the strained material did not contribute to energy dissipation. This radius was only roughly estimated, and hence Mason's value of B must be considered to be only an order of magnitude calculation.

It is generally believed that in the liquid helium temperature range, conduction electrons provide a significant contribution to energy absorption from a uniformly moving dislocation in a continuous lattice. However Brailsford[25] has demonstrated that conduction electrons interact strongly only with the dilatational field of moving dislocations. Hence on the basis of isotropic elasticity theory, only the edge component of a moving dislocation is influenced by the electron gas. Brailsford's edge dislocation damping coefficient is calculated to be 0.25×10^{-5} cgs for copper, 0.39×10^{-5} cgs for lead, and 0.76×10^{-5} cgs for aluminum. These theoretical values are smaller by about a factor of two than the total damping coefficient obtained experimentally for both lead[31] and aluminum[32] in internal friction measurements at 4.2°K. (The effective dislocation Burgers vector operating in the internal friction experiments is undetermined. In the case where this Burgers vector is not pure edge, then the applicable theoretical values for the damping coefficient are less than those given above, thereby increasing the difference between theoretical and experimentally determined results.) For copper, the damping coefficient for dislocations of 30° mixed character was determined experimentally to be 0.83×10^{-5} cgs.[+] Brailsford's theoretical calculation for the electronic damping coefficient for copper, applied to the edge component of 30° mixed dislocations only, is equal to 0.06×10^{-5} cgs, a factor of thirteen smaller. It is as yet not clear whether closer agreement between experiment and theory can be obtained by further improvement of the experimental work or by extension of the theoretical analysis to include such effects as anisotropic elasticity.

Results of direct mobility experiments by Weertman and coworkers[28,29] indicate that below 30°K, B in aluminum and in the

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[+] To be reported elsewhere.

normal state in lead rises dramatically, in stark contrast to the results discussed above. The discrepancy between these and those observations discussed above has not yet been explained although it is noted that in the most recently published work in aluminum,[29] B is estimated by averaging data points, rather than using the data of the maximum asymptote as is now commonly accepted.

The unresolved aspects of intrinsic dislocation interactions suggest work in the following areas:

(i) theoretical treatment of dislocation-phonon interactions which account for crystal anisotropy. This is especially important in the case of phonon viscosity where large errors may be introduced by the use of an averaged Gruneisen number rather than directionally dependent values.

(ii) numerical calculation of the damping coefficient derived from discrete lattice effects (phonon dispersion relation) at low temperatures.

(iii) experimental measurement of the damping coefficient of screw dislocations at $4.2^{o}K$. Conduction electrons do not interact significantly with screw dislocations because of the absence of a dilatational stress field (in the case of elastic isotropy) and hence residual damping at $4.2^{o}K$ must arise from other sources.

(iv) further direct measurement of dislocation mobility in the normal and superconducting states of superconducting metals. The difference in energy dissipation between the two states provides a direct measure of conduction electron damping.

FIGURE CAPTIONS

Fig. 1. Basal surface of a zinc test crystal after application of a torsional stress pulse showing dislocation displacements from both continuous (at right) and discontinuous scratches. The cylindrical axis of the crystal (and position of zero shear stress) is at the bottom center. The applied resolved shear stress did not exceed the breakaway stress for the discontinuous scratch segment closest to the cylindrical axis and its dislocation loops collapsed when the stress was removed.

Fig. 2. Basal cleavage surface of zinc with a 0.63 mm scratch after application of a 10 μsec stress pulse at 66^{o}K.

Fig. 3. Damping coefficient as a function of temperature for dislocations in copper.

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Figure 1. Basal surface of a zinc test crystal after application of a torsional stress pulse showing dislocation displacements from both continuous (at right) and discontinuous scratches. The cylindrical axis of the crystal (and position of zero shear stress) is at the bottom center. The applied resolved shear stress did not exceed the breakaway stress for the discontinuous scratch segment closest to the cylindrical axis and its dislocation loops collapsed when the stress was removed.

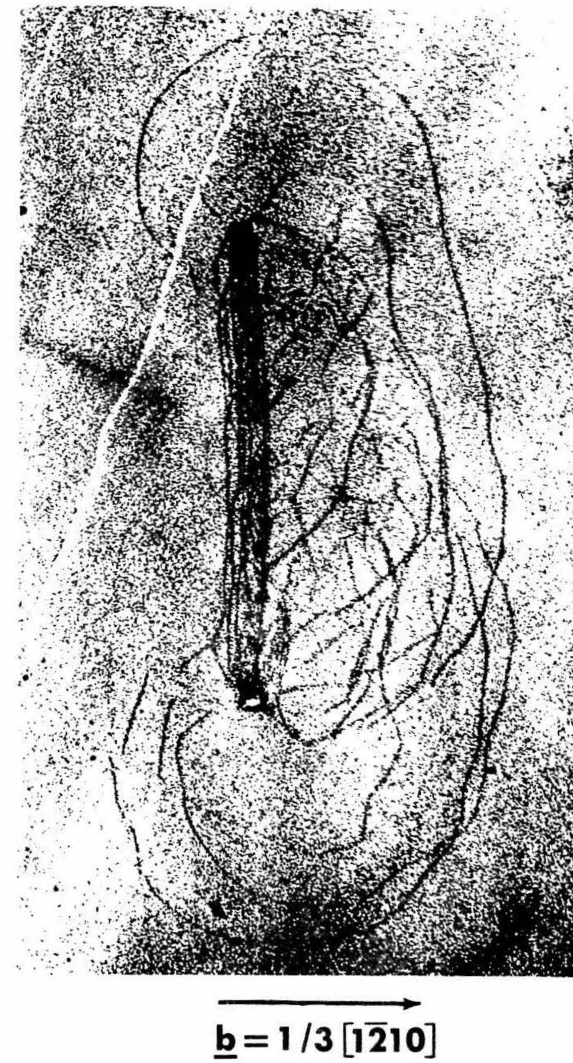

Figure 2. Basal cleavage surface of zinc with a 0.63 mm scratch after application of a 10 μsec stress pulse at 66°K.

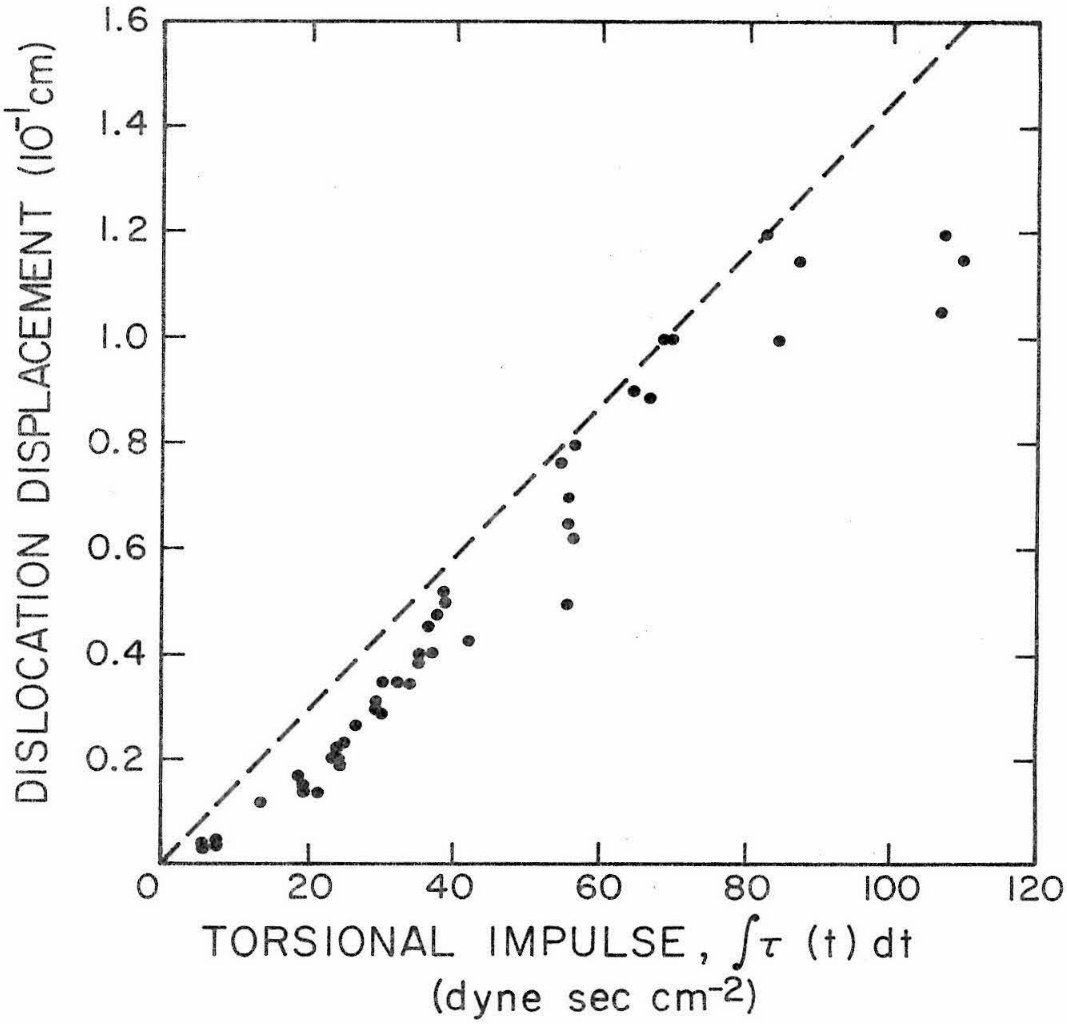

Figure 3. Dislocation displacement as a function of total torsional impulse, for 30° mixed dislocations in copper at 4.2°K.